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MECHANICAL PROPERTIES AND ABRASION RESISTANCE OF EXPERIMENTAL DIRECT-QUENCHED STEELS

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MECHANICAL PROPERTIES AND ABRASION RESISTANCE OF EXPERIMENTAL DIRECT-QUENCHED STEELS

by

R.F. Knight*

ABSTRACT

The microstructures, mechanical properties and abrasion resistance of several control-rolled, direct-quenched steels were investigated. Materials studied included four C-Mn-Mo-Nb steels, five steels with compositions based on CHT 360, and three commercial steels.

Hardness levels and abrasion resistance equivalent to that of commercial CHT 360 were developed in some of the direct-quenched alloys. However, no advantage attributable to direct-quenching was apparent. Significant reductions in carbon content could not be made with equivalent abrasion resistance still being achieved. In addition, anisotropy introduced by rolling to low temperatures resulted in poor transverse tensile ductility and impact toughness.

In low stress abrasion tests, abrasion resistance was hardness dependent. High strength abrasion resistance was independent of hardness level in the range studied.

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LABORATOIRES DE RECHERCHE EN MÉTALLURGIE PHYSIQUE
RAPPORT MRP/PMRL 83-75(TR)

PROPRIÉTÉS MÉCANIQUES ET RÉSISTANCE À L'USURE D'ACIERS EXPÉRIMENTAUX TREMPÉS DIRECTEMENT

par

R.F. Knight*

RÉSUMÉ

On a étudié les microstructures, les propriétés mécaniques et la résistance à l'usure de plusieurs aciers trempés directement et laminés en conditions contrôlées. Les produits étudiés comprenaient quatre aciers au C-Mn-Mo-Nb, cinq aciers à compositions basées sur le CHT 360 et trois aciers commerciaux.

On a réussi à produire dans certains des aciers trempés directement des niveaux de dureté et une résistance à l'usure comparables à ceux du CHT 360. Cependant, il n'y avait aucun avantage évident attribuable à la trempe directe. Il était impossible de réduire de façon appréciable la teneur en carbone sans que la résistance à l'usure soit réduite. De plus, l'anisotropie créée par le laminage à basses température s'est traduite par une ductilité en traction transversable et une résilience au choc faibles.

Dans les essais d'usure par frottement sous faible contrainte, la résistance à l'usure dépendait de la dureté. Dans la plage étudiée, la résistance à l'usure sous contrainte élevée était indépendante du niveau de dureté.

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INTRODUCTION

Earlier research on linepipe steels (1,2) indicated that controlled rolling and direct quenching can result in a good combination of strength and toughness. There were indications that higher strength was attainable than by conventional reaustenitizing and quenching. For low alloy steels, abrasion resistance increases with increased hardness, and hence, strength (2). It was argued that direct quenching of "higher-carbon" microalloyed steels might result in superior hardness and abrasion resistance, or, better ductility and toughness for a given abrasion resistance.

A series of C-Mn-Mo-Nb steels with varying carbon contents (M1 to M4), were prepared together with a series of steels based on the commercial CHT 360 composition, also with varying carbon contents and with and without additions or niobium (E1 to E5). These steels were processed by controlled rolling and direct quenching, and their mechanical properties, microstructures, and abrasion-resistance determined. Comparisons were made with the properties of three commercial steels, CHT 100 (C1), CHT 360 (C2) and CHT 475 (C3), in the as-received (quenched and tempered) condition.

EXPERIMENTAL PROCEDURES

MATERIALS

The steels were induction-melted under an argon cover. The charge material was electrolytic iron for the M steels and commercial billet material for the E steels. High-purity ferroalloys were used for additions. The chemical compositions, determined by quantitative spectrography, as well as those for the commercial steels, are shown in Table 1.

PREPARATION OF ROLLING STOCK

The 45 kg M steel ingots were cropped and sliced in half longitudinally. These slabs were soaked at 1230°C and rolled to 13 mm plate, with one half direct-quenched (DQ) and the other half air-cooled (AC).

The E steels were poured into tapered, sand-moulded billets, soaked at 1095°C, and rolled to 57 mm thick slabs. The slabs were cut in two to facilitate handling and soaked at 1200°C, and rolled into 13 mm plate.

ROLLING

All steels were rolled to 13 mm by the schedule shown in Table 2. However, because of their higher alloy content, the E steels required an extra pass between 900°C and 800°C, compared with the M steels.

COOLING OF ROLLED PLATE

All experimental compositions except E5 were processed to obtain the DQ condition by water-spray quenching immediately after the final rolling pass. The mean cooling rate between 750°C and 500°C, as monitored by a thermocouple at the mid-length/mid-thickness position, was 85°C per second.

Steel E5 and the opposite halves of M1, M2 and M4 were air-cooled after the final pass to obtain the AC condition.

HEAT TREATMENT

Samples of the M steels in the AC and DQ conditions were subjected to heat treatment tests (3) to determine suitable tempering treatments to give a hardness equal to that of CHT 360 (340 to 380 BHN). Hardness was determined by averaging numerous Rockwell "A" readings on ground through-thickness sections. The results, some of which are summarized in Table 3, indicated that suitable tempering treatments for the DQ steels would be 1 h at 315°C for M2 and M3 and 1 h at 370°C for M4. These treatments were applied to the plates to be tested. Steel M1 (low carbon), which did not develop the required hardness, was tested in the DQ condition.

Steels E1, E2, E3, and E4 were tested in the DQ conditions. Steel E3 was also tested after tempering for 1 h at 300°C. Steel E5 was reaustenitized and quenched and then tempered for 1 h at 370°C.

TESTING

a - Microstructure

Photomicrographs of all the test materials (longitudinal sections, etched with 2% nital) were obtained.

b - Hardness

The quarter-thickness hardness of all the experimental materials was determined by averaging a minimum of 10 Rockwell "C" readings on ground, longitudinal cross-sections.

c - Mechanical properties

Duplicate tensile specimens were tested for all experimental materials. The specimens had 4.5 mm gauge diameters and were machined at the mid-thickness of the plate. Charpy V-notch impact specimens, notched through the plate thickness, were tested at various temperatures to establish the impact characteristics. Properties transverse to the rolling direction were established for all steels; longitudinal properties were also determined for the E steels.

d - Abrasion resistance

Abrasion-resistance testing was performed by Falconbridge Limited under contract (4). Operational problems prevented the preparation of all steels in time to be tested. The steels tested were: M1 (DQ); M2 (DQ + 315); M3 (DQ + 315); M4 (DQ + 370); E1 (DQ); E3 (DQ + 300); E5 (AC + Q + 370). Three commercial steels (CHT) were also tested.

Two test procedures were employed to simulate low-stress and high-stress conditions. The abrasive used was AFS 50/70 subangular silica. Figure 1 shows the test specimen shape used.

Resistance to low-stress abrasion, such as that encountered by conveyors for bucket-wheel excavators, truck boxes, etc., was examined using the "rubber wheel test" described by ASTM G 65-80 (Procedure B). This is illustrated by Fig. 2. In this test, light pivotal loading forces a flat specimen against a rotating rubber wheel which drags gravity-fed abrasive across the surface. The abrasive slides along the surface and is not comminuted.

High-stress abrasion tests were carried out with apparatus patterned on that developed originally by BCIRA, as illustrated in Fig. 3. In this test a 30.6 kg load is applied to a flat sample which is moved along a sand-covered steel track. The abrasive is crushed between the specimen and the track and, hence, has to be replaced regularly. This type of abrasive action is similar to that occurring in jaw crushers and hammer mills.

The low-stress test procedure required testing four samples of each steel, with each sample tested for two 10-min periods. The high-stress test required testing of four samples of each steel for three 10-min periods, the sand being replaced after every group of three tests. Every fourth sample tested was a mild steel reference sample.

Abrasion resistance is expressed as an "abrasion factor":

$$AF = \frac{\text{Av. weight loss of reference sample}}{\text{Weight loss of test sample}}$$

RESULTS

Figure 4 shows the microstructures of all the experimental steels in the conditions in which they were tested for abrasion resistance. The tensile properties are summarized in Table 4. Figure 5 shows the transverse impact properties and Fig. 6 shows the longitudinal impact properties of the E steels. The hardness and abrasion factors (with their 95% confidence intervals) are shown in Table 5 and Fig. 7.

DISCUSSION

Figure 4 shows that all experimental direct-quenched steels had elongated grain structures promoted by controlled rolling to low finishing temperatures. The transformation product was acicular in nature. Steel E5, which was reaustenitized and quenched, had a more uniform microstructure similar to that of the quenched and tempered commercial steels. The use of electrolytic iron as melting stock for the M steels resulted in a lower sulphur content and fewer nonmetallic inclusions than were found in the other steels.

Table 3 shows that a higher hardness can be achieved from direct quenching than by reaustenitizing and quenching and that this hardness increment is retained after tempering. However, in the CHT 360 type steels, the small hardness increment secured in this way is not enough to offset the loss of hardness that attends a significant reduction in carbon content.

The E steels were tested in orientations longitudinal and transverse to the rolling direction. These steels showed a high degree of anisotropy which takes the form of decreased transverse ductility (Table 4) and transverse toughness (Fig. 5 vs Fig. 6). Steel E5 also showed high anisotropy in spite of the fact that it was reaustenitized and quenched indicating that the anisotropy is attributable to the effects of rolling to low temperature rather than to the quenching operation.

Comparison of the tensile and impact properties of steel E3 in the quenched and quenched-and-tempered conditions indicates that tempering had a detrimental effect (Fig. 6 and Table 4). Tempering at higher temperatures would probably result in improved ductility and toughness, but it would also result in the hardness being decreased to a level of little interest for abrasive applications.

In all cases the M steels had better transverse impact properties than the E steels (Fig. 5). All experimental steels with hardnesses near the target range had poorer impact properties than a sample of CHT 360. Figure 6 shows that steels E1 (low strength) and E3 had better longitudinal impact properties than values typical of commercial CHT 360 as indicated by circles (6). Steels E2 and E5 were roughly equivalent to CHT 360 whereas steel E4, the hardest material, was inferior in toughness.

Figure 7 shows that, for the steels tested, low stress abrasion resistance increases with hardness whereas high stress abrasion resistance is independent of hardness. The values reported were in good agreement with the abrasion resistance of other low alloy steels tested in a similar manner (4). Low stress abrasion was characterized by a smooth, grooved appearance of the abraded surface, as shown in Fig. 8. This type of abrasion appears to result in a built-up edge on the abrasive material and a uniform shearing of metal chips from the abraded surface. The abraded surface from the high stress tests has a roughened appearance, as shown in Fig. 9. This appearance results from the impression of abrasive particles into the sample surface, impact fragmentation of the particles to form sharp edges and irregular chipping-off of the surface layer of the sample. There was no significant difference in the appearance of the abraded surfaces as a function of hardness.

The high abrasion resistance characteristic of hard-faced materials are unattainable with low alloy steels. For example, a highly alloyed white iron with a hardfacing deposit at Rc 68 reportedly has abrasion factors of 30 $\pm$ 10 for low stress and 3.0 $\pm$ 0.3 for high stress (4). None of the experimental steels showed abrasion factors higher than would be expected of low alloy steels of the hardness tested. Nevertheless, abrasion factors were achieved by direct quenching which are equivalent to those of commercial CHT 360 (Sample C2), which has been demonstrated to be a useful material for abrasion resistant applications.

The abrasion factors reported are based on "mild steel" reference samples. Softer reference material would result in higher absolute numbers.

CONCLUSIONS

Direct quenched materials can develop the hardness and abrasion resistance of commercial abrasion resistant steels.

There is no particular advantage attributable to direct quenching in terms of abrasion resistance. There are indications of higher hardness levels being attainable than those obtained by reaustenitizing and quenching techniques. However, the hardness increment is small and is of no practical value because it does not allow the use of significantly lower carbon contents to achieve equivalent hardness.

The experimental steels developed a high degree of anisotropy as a result of rolling at low temperatures. This produced poor tensile ductility and impact toughness in an orientation transverse to the rolling direction.

The low stress abrasion resistance of all the materials tested increased as their hardness increased. Their abrasion factors showed no significant difference from those of other low alloy steels. At the hardness level required for improved abrasion resistance, tempering of direct quenched steel did not enhance its ductility or toughness.

The hardness independence of high stress abrasions factors indicates that for high stress applications little is to be gained by using low alloy steels at high hardness levels. Significant improvements for such applications can only be achieved at much higher hardness levels, attainable by techniques such as hardfacing.

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Table 1. Chemical composition (wt %).

Steel	C	Mn	Si	S	P	Al	Mo	Nb	V	Ti	B
M1	0.10	1.50	0.29	0.011	0.006	0.08	0.24	0.065	-	-	-
M2	0.16	1.48	0.24	0.015	0.010	0.04	0.20	0.060	-	-	-
M3	0.17	1.34	0.21	0.010	0.010	0.03	0.21	0.060	-	-	-
M4	0.19	1.45	0.21	0.010	0.010	0.03	0.20	0.065	-	-	-
E1	0.04	1.36	0.23	0.037	0.008	0.07	0.24	0.060	0.05	0.05	0.0020
E2	0.12	1.37	0.25	0.035	0.008	0.08	0.24	0.052	0.04	0.05	0.0015
E3	0.15	1.32	0.26	0.035	0.009	0.02	0.23	0.055	0.04	0.02	0.0020
E4	0.19	1.35	0.33	0.030	0.007	0.05	0.26	-	0.04	0.04	0.0040
E5	0.18	1.41	0.27	0.035	0.010	0.04	0.25	-	0.04	0.06	0.0019
C1	0.13	1.42	0.29	0.014	0.007	0.07	0.21	-	<.005	0.04	<.0001
C2	0.13	1.42	0.29	0.020	0.004	0.08	0.27	-	<.005	0.04	0.0003
C3	0.28	1.39	0.29	0.014	0.003	0.07	0.57	-	<.005	0.04	0.0004

Table 2. Rolling schedules.

Operation	M Steels		E Steels	
	Gap (mm)	T (°C)	Gap (mm)	T (°C)
Soak	-	1230	-	1200
Pass 1	50.8	1205	50.0	1175
Pass 2	44.4	to	43.7	to
Pass 3	40.3	1165	39.6	1125
Pass 4	36.4	900	35.8	900
Pass 5	30.5	875	30.5	880
Pass 6	22.6	850	25.4	860
Pass 7	15.7	825	20.3	840
Pass 8	10.7*	800	15.2	820
Pass 9	-	-	10.7*	800

* - 13 mm after spring-back.

Table 3. Hardness of heat treatment samples*.

Condition	M1	M2	M3	M4
AC	185	195		205
AC + Q	238	305		322
AC + Q + 370 (1 h)	235	283		309
DQ	276	337	377	393
DQ + 315 (1 h)	268	357	362	372
DQ + 370 (1 h)	258	309	305	318
DQ + 480 (1 h)	250	276	290	294

*BHN converted from R_A.

Table 4. Tensile properties.

Steel	Condition	Orientation	UTS (MPa)	YS (MPA)	El. (%)	RA (%)
M1	DQ	T	1007	812	15.0	44.0
M2	DQ + 315	T	1249	1119	7.7	23.0
M3	DQ + 315	T	1256	1168	6.0	17.0
M4	DQ + 370	T	1229	1158	7.7	23.7
E1	DQ	T	787	640	18.0	35.0
"	"	L	747	643	24.5	64.5
E2	DQ	T	1170	1026	11.0	22.5
"	"	L	1124	979	17.5	51.0
E3	DQ	T	1145	1000	10.0	19.5
"	W	L	1121	963	20.5	51.0
"	DQ + 300	T	1096	986	7.5	14.5
"	"	L	1037	937	17.5	53.0
E4	DQ	T	1531	1383	8.0	19.5
"	DQ	L	1404	1262	16.0	47.5
E5	AC + Q + 370	T	1246	1153	10.5	27.0
"	"	L	1236	1153	16.0	55.5
C2	As rec.	T	1279	1216	12.3	50.0
CHT 360	Typical	L	1240	1138	11	30

Table 5. Abrasion factors (AF) for plate steels.

Steel	Condition	BHN*	Low-stress AF	High-stress AF
M1	DQ	276	1.15 ± .03	1.28 ± .06
M2	DQ + 315	362	1.31 ± .06	1.30 ± .05
M3	DQ + 315	362	1.31 ± .03	1.29 ± .06
M4	DQ + 370	404	1.73 ± .06	1.2 ± .1
E1	DQ	243	1.01 ± .02	1.14 ± .08
E3	DQ + 300	337	1.3 ±.06	1.32 ± .07
E5	AC + Q + 370	382	1.50 ± .05	1.30 ± .09
C1	WQ + 650	238	1.05 ± .05	1.1 ± .1
C2	WQ + 370	377	1.51 ± .06	1.37 ± .04
C3	WQ	503	2.4 ± .2	1.2 ± .1

* - Av quarter-thickness hardness of abrasion samples (Rc converted to BHN).

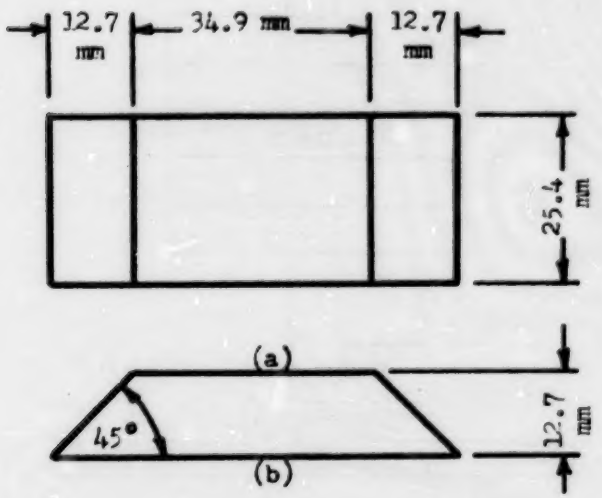

Fig. 1. Configuration and approximate dimensions of abrasion test specimen. (a) - high stress; (b) low stress.

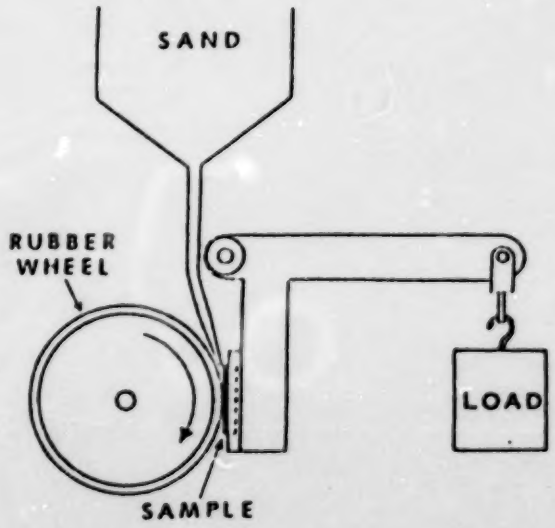

Fig. 2. Low stress abrasion test.

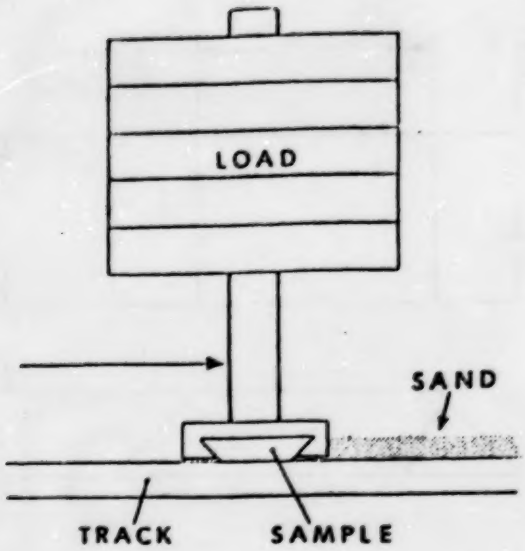

Fig. 3. High stress abrasion test.

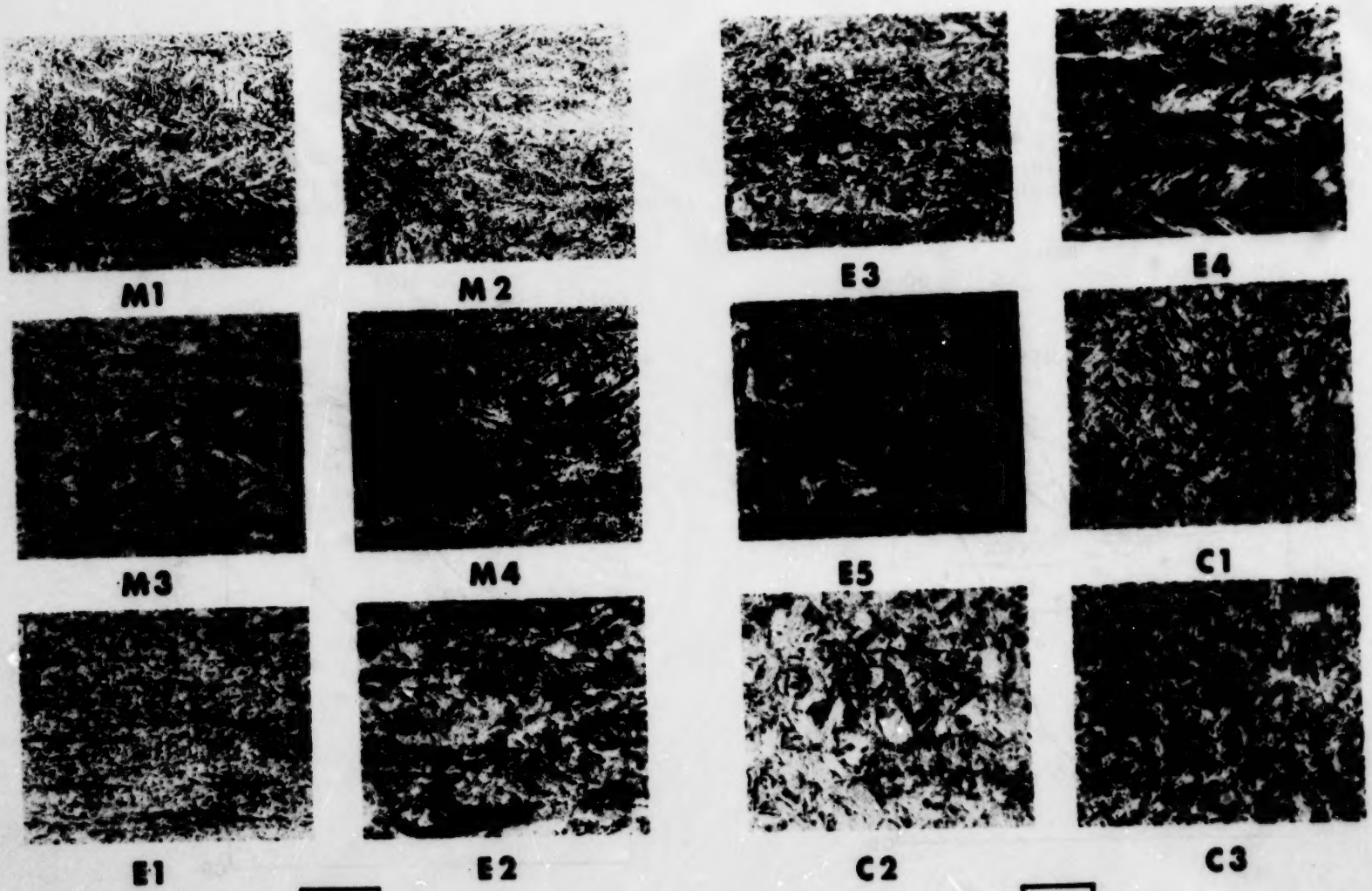

Fig. 4. Microstructures of experimental steels (Etched in 2% Nital).

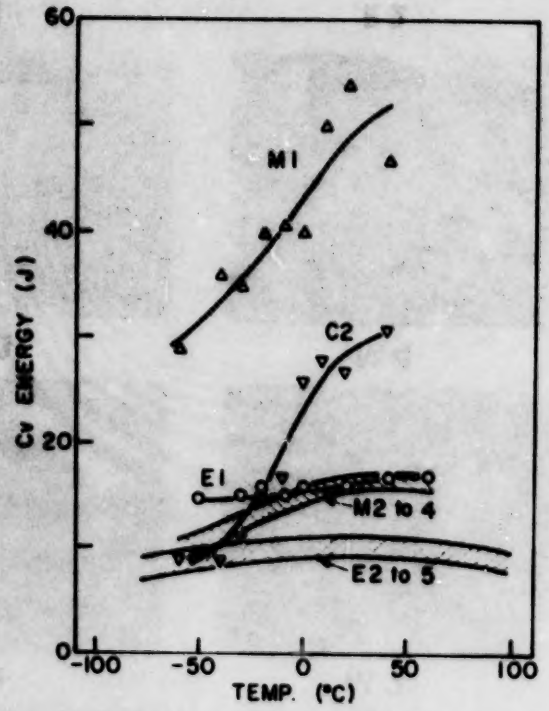

Fig. 5. Transverse impact properties of experimental steels. Steel C2 is a sample of CHT 360 other than that tested for abrasion. (neg 9279)

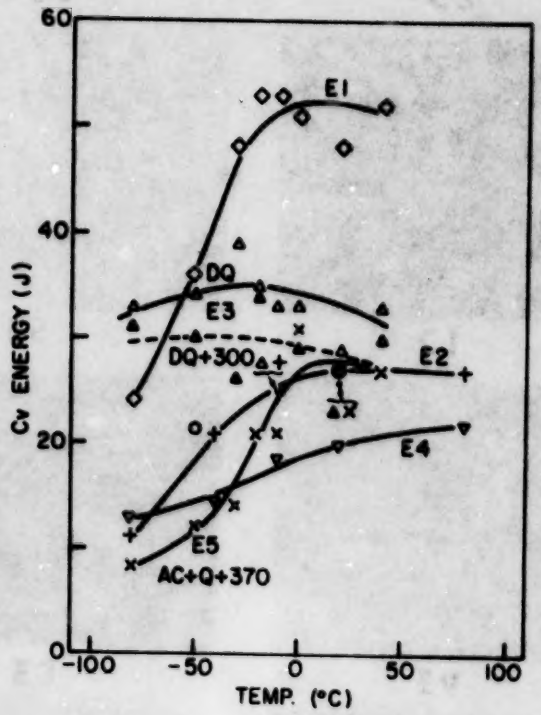

Fig. 6. Longitudinal impact properties of E steels. Typical values for CHT 360 are indicated by circles. (neg 9278)

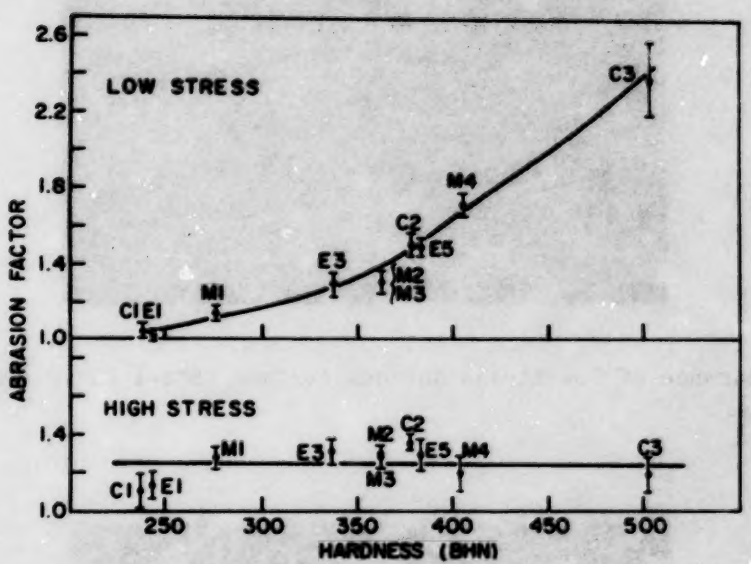

Fig. 7. Effect of hardness on abrasion factors of experimental steels. (Neg. 9277)

Fig. 8. Appearance of low stress abraded surface (Steel C2). (neg 8803)

Fig. 9. Appearance of high stress abraded surface (Steel C2). (neg 8801)